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AFML USAF ltr, 12 Jan 1972

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review

OF RECENT
DEVELOPMENTS

LIQUID METALS

E. M. Simons • July 15, 1970

CORROSION BY SODIUM

Corrosion of Stainless Steels by Sodium

An attempt was made at Brookhaven to define the corrosion mechanism of stainless steels by studying the corrosion and interaction of pure cobalt, iron, nickel, and chromium in sodium. (1) A continuous string of polished specimens arranged in repeated groups of four in the order listed was exposed for 1000 hours to 1200 to 1400 F sodium in the hot leg of a Type 316 stainless steel thermal-convection loop (about 0.1 ft/sec sodium velocity). The diffusion-type cold trap was maintained at 225, 390, or 700 F, giving measured oxygen levels in the sodium of 4, 10, or 110 ppm, respectively. The following observations resulted:

- (1) All specimens were etched by grain-boundary attack except for the chromium specimens exposed to 110-ppm-oxygen sodium. These showed a dark gray surface film.
- (2) Nickel specimens lost weight (up to 2 mg/cm²), while iron and cobalt specimens gained weight (up to 0.8 mg/cm²). Chromium specimens showed little or no weight change.
- (3) All four elements migrated. As the oxygen content of the sodium was increased from 4 to 110 ppm, deposition of iron increased sharply, deposition of nickel did not change, deposition of chromium increased slightly, and only traces of cobalt deposition were found.

In order to explore the limits of performance at potential hot-spot conditions in a liquid-metal-cooled, fast-breeder reactor (LMFBR), General Electric extended its previous 1200 F maximum-temperature loop studies to 1300 F. (2) Corrosion specimens of Types 304 and 316 stainless steel were exposed for periods up to 3000 hours to sodium containing less than 10 ppm oxygen and flowing at up to 30 ft/sec with a cold-leg temperature of 800 F. The researchers summarized their findings as follows.

- (1) Corrosion rates at sodium velocities less than about 15 ft/sec are approximately linearly dependent on velocity. Increasing sodium velocity above 15 ft/sec produces a progressively smaller

increase in corrosion until 30 ft/sec, above which corrosion appears independent of velocity.

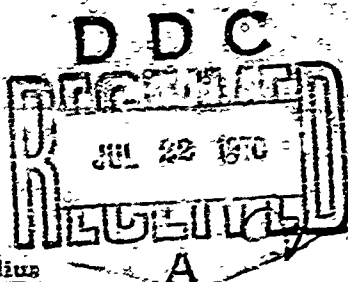
- (2) Corrosion rates measured at a single velocity but at successive positions downstream (with resultant increasing saturation of the sodium with corrosion products) were found to decrease by 50 to 75 percent.
- (3) Changing sodium mass flow in the range 0.5 to 1.0 gpm at the same sodium velocity (about 3 ft/sec with Reynold's number less than 10,000) produced less than 10 percent increase in corrosion rate.
- (4) Significant reductions in chromium, nickel, and manganese were observed on surfaces of corrosion samples. The relative amount of loss for each was dependent on exposure conditions. In separate experiments that differed in sodium velocity and, therefore, in total corrosion, the pattern of movement of chromium and nickel was different but manganese stayed the same. Nickel movement was more sensitive to velocity than chromium, showing a greater loss at higher sodium velocities.
- (5) An experiment comparing corrosion of pure iron surfaces to corrosion of Type 316 stainless steel surfaces in a loop hot leg showed that alloy composition played a significant role in the corrosion process. The iron surfaces gained weight and approached a composition near that of the stainless steel surfaces.
- (6) Turbulence and activity of alloy constituents in sodium solution played the dominant role in cold-leg deposition. An increase of sodium velocity from 5 to 10 ft/sec resulted in 85 percent increase in rate of deposition of corrosion products.
- (7) Various carbides (M₂₃C₆, M₇C₃) and NaCrO₂ have been identified in the deposits from the cold leg.

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- (8) Nucleation of particulates in the sodium stream upon cooling and subsequent deposition on cold-leg surfaces is believed to be a more likely process than diffusion of elemental species to surfaces followed by nucleation.
- (9) Contrary to equilibrium considerations, carbon transport was observed from high to lower temperature regions. A correlation was found in two separate test runs between movement of carbon and chromium. Carbon loss as high as 79 percent was measured for Type 316 stainless steel at 1300° F.
- (10) Microstructural changes were observed in all alloys and appeared to be dependent on corrosion rate and alloy composition. Three general changes were seen: (a) ferric surface phase, (b) a zone adjacent to the surface and depleted in carbon, and (c) localized and enhanced elemental leaching at grain boundaries.
- (11) Carbon movement had a profound effect on microstructural changes, by reducing the normal carbide precipitation and enhancing formation of intermetallic phases. Continued development of these changes implies major effects on long-term creep and rupture properties.

Loop experiments were conducted at Atomic International in which high-purity 1200 F sodium was saturated with corrosion products at the inlet to the Type 316 stainless steel test section, while the test-section surface was subjected to a constant heat flux of about 10^6 Btu/(ft²)(hr) by internal heating to produce a 55 F temperature rise along its 3.5-inch length.⁽³⁾ The objective was to note the effects of velocity (10 and 20 ft/sec) and temperature on corrosion without introducing extraneous system-dependent effects. From the observed velocity effects, the corrosion rate was concluded to be controlled by liquid diffusion.

With the difference between the concentration of dissolved metal in sodium next to the surface and of the dissolved metal in the bulk of the liquid as the driving force for metal dissolution, expressions were derived for the maximum corrosion rates at various locations along the test section and then compared with the measured rates. Reasonable agreement was obtained when the lowest published solubility data of iron in sodium was used, indicating that the theory may be useful for predicting the maximum corrosion rate of cladding under reactor operating conditions.

Effects of Sodium on Mechanical Properties on Component Materials

Loop tests were conducted at Atomic International to evaluate the effects of steady and cyclic heat fluxes [0.8 and 1.0×10^6 Btu/(ft²)(hr)] on creep behavior of Types 304 and 316 stainless steel tubes immersed in high-purity 1200 F sodium flowing at 20 ft/sec.⁽⁴⁾ Mass transfer was observed under steady internal heating for up to 600 hours, but there was no difference in creep behavior from that of tubes stress-rupture tested in static sodium. Cyclic heating reduced the rupture life by a factor of 2 and produced large changes in microstructure, strain profile, and strain at failure. These effects were rela-

tively insensitive to the magnitude of the cyclic heat flux. However, an increase in frequency reduced the life of both annealed and cold-worked Type 304 stainless steel and had the opposite effect on cold-worked Type 316 stainless steel.

The analysis indicates that repeated thermal-stress relaxation, as a result of heat-flux cycling, produces an increment of plastic strain for each cycle, superimposed on conventional biaxial creep due to pressure. Though this study is inadequate to establish design criteria for cyclic action, it does point to danger in designing fuel sheaths on the basis of steady-state behavior. The numerous reactor transients that will be experienced during service could be sufficient to produce end-of-life ductility limits of irradiated cladding.

Mine Safety conducted an extended study on the effects of sodium and helium on design properties of Types 304 and 316 stainless steel to see whether the ASTM allowable-code values for stainless steels in air could be extended to cover the design of an LMFBR system. Test conditions included stress-free exposures to high-oxygen sodium, high-carbon sodium, clean sodium, and helium at 1200 and 1400 F; tensile tests in air at 1200 and 1400 F; and creep-rupture tests at 1200 F in flowing (20 ft/sec) clean sodium for times up to 4000 hours. The summary report concluded that, on the basis of the tensile data and extrapolated creep data, Type 304 stainless steel code values could be applied to an LMFBR design with allowances being made for the anticipated losses in ductility.⁽⁵⁾ For Type 316 stainless steel, however, some reduction in code values would be warranted, with attention still given to the ductility losses noted.

The investigators recommended that, for both steels, long-term (greater than 10,000-hour) creep data be obtained to determine the true longer term creep behavior instead of relying on extrapolation of 4000-hour data. The Type 316 stainless steel appeared stronger than the Type 304 in most cases during short-time testing. However, when the data were extrapolated, the strength differences were less apparent. In fact, in some cases, the Type 304 alloy appeared stronger. Of course, the fact that the results are based on single heats of material and long extrapolations suggests that they should not be considered final.

The following observations were made on the basis of measurements of the room-temperature and 1200 F mechanical properties of a single heat of Type 316 stainless steel during exposure to air, helium, and low-oxygen sodium:⁽⁶⁾

Test	Environment		
	Air	Helium	Sodium
Cyclic strain	Fewest cycles to failure	Most cycles to failure	Intermediate
Stress rupture	No significant difference in rupture strength; ductility perhaps better in sodium		
Creep	Lowest rate	Highest rate	Intermediate
Tensile strength	Small differences (less than 6 percent)		

When the quantitative data from these tests were compared with ASM Metals Handbook data from several heats of Type 316 stainless steel tested in air, the variations in properties owing to environmental differences were noted to be substantially less than the variations owing to differences from heat to heat of the steel. This is apparent from the following tabulation:

	MSAR	ASM Data
Heats of Type 316 Stainless Steel	1	13
Exposure Environments	Air, He, Na	Air
Temperature	1200 F	1200 F
Ultimate Tensile Strength Range, ksi	48-51	42-64
0.2% Offset Yield Strength Range, ksi	26-27	12-39
Elongation Range, percent	44-48	23-63

A study of the application of Section VIII of the ASME Boiler and Pressure Vessel Code to the determination of allowable design stresses for low-alloy steels in bimetallic sodium systems was conducted at General Electric.⁽⁷⁾ The decarburization that occurs during operation of such systems, coupled with the corrosion and creep, leads to progressively lower material strength and higher stresses. The initial design features of a model system were determined for three different cases under conditions at which the strain rate was the controlling parameter. The rate of decarburization to the equilibrium level influences the results. Since this is controlled by geometry and fluid dynamics, presence of other alloys, temperature profile, sodium chemistry, and other factors, calculations were made for both rapid and slow decarburization.

Table 1 shows the results for each case. The initial design stress for an expected lifetime of 100,000 hours depends on the case, and ranges from 69 to 78 percent of the code-allowable stress. Note that the thin wall in Case I leads to end-of-life strains and strain rates that would probably exceed the capabilities of the alloy. Case II leads to end-of-life strain rates that are higher than the code-allowable value of 0.01%/1000 hours. The use of total accumulated strain as the controlling criterion (Case III), while requiring the greatest wall thickness, appears to be the most practical choice.

Corrosion of Vanadium Alloys by Sodium

As part of an Oak Ridge program to study interstitial mass transport between stainless steel and vanadium, 40-mil sheet vanadium tensile specimens are being exposed for 500 hours to 700 and 800 C (1290 and 1470 F) low-oxygen sodium in stainless steel containers.⁽⁸⁾ Two grades of vanadium, commercial and high purity, were tested. The high-purity material increased in oxygen, carbon, and nitrogen, whereas the commercial material gained only carbon and nitrogen. The transport of carbon and nitrogen from Type 304L stainless steel to the vanadium was substantially greater at 800 C than at 700 C. Interstitial transport from titanium-stabilized Type 321 stainless steel was negligible at 700 C compared with that from Type 304L stainless steel. Room-temperature tensile ductilities of both grades of vanadium were reduced by the sodium exposures in Type 304L stainless steel containers. In contrast, testing in Type 321 stainless steel produced no measurable change in either strength or ductility compared with the properties of the control specimens after exposure to 700 C argon for 500 hours.

In similar studies at Westinghouse with vanadium-alloy inserts in pumped low-oxygen sodium loops made of Type 316 stainless steel tubing and operating between 1250 and 1470 F, the inserts gained weight as a result of major increases in nitrogen, carbon, and oxygen, particularly near the exposed surfaces.⁽⁹⁾ For example, after 1500 hours in 1450 F sodium containing less than 10 ppm oxygen

TABLE 1. SUMMARY OF DESIGN FEATURES AND OPERATING BEHAVIOR OF LOW-ALLOY STEEL FOR DIFFERENT INTERPRETATIONS OF ASME CODE⁽⁹⁾

Model System						
Temperature = 900 F						
Initial OD of Tubing = 1.50 in.						
Pressure = 1800 psi						
Expected Lifetime = 100,000 hours						
Corrosion rate of steel in sodium = 0.003 in./100,000 hours.						
ASME Code Interpretation	Initial Wall Thickness, in.	Initial Design Stress, psi	Estimated Operating Stress, psi	End of Life		ASME Code
				Total Strain, percent	Strain Rate, %/1000 hours	
<u>Slower Rate of Decarburization</u>						
Case I (a)	0.106	13,100	15,990	10.3	0.32	
Case II (b)	0.133	10,150	10,650	1.0	0.032	
Case III (c)	0.149	9,060	9,300	0.3	0.01	
<u>More Rapid Rate of Decarburization</u>						
Case I (a)	0.106	13,100	19,080	20.3	0.33	
Case II (b)	0.142	9,510	9,910	1.0	0.0175	
Case III (c)	0.149	9,060	9,370	0.65	0.01	

- (a) Initial strain rate at startup = 0.01%/1000 hours.
(b) Total creep strain after 100,000 hours = 1 percent.
(c) Strain rate at end of life = 0.01%/1000 hours.

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and flowing at 5 ft/sec, the 0.5- μ il surface layer of a Vanstar-9 sample (V-6.4Fe-5.3Cb-1.3Zr-0.06C) showed the following increases in interstitial content: nitrogen went from 89 to 21,000 ppm; carbon went from 500 to 5500 ppm; and oxygen went from 550 to over 6000 ppm. The Type 316 stainless steel containment tubing, on the other hand, showed severe depletion of both nitrogen and carbon at the sodium-exposed surface and nitrogen enrichment at the air-exposed surface. Using available interstitial diffusion data, the researchers calculated the theoretical weight loss of interstitials in the stainless steel under test-loop conditions and got excellent agreement with the measured weight gain in the vanadium alloy. This provides a method of predicting the effects of material and surface-area changes on the behavior of vanadium alloys in sodium.

At Argonne, vanadium and selected higher strength vanadium-base alloys were exposed to sodium cold trapped at two temperatures (about 250 and 350 F, corresponding to about 6 and 15 ppm oxygen, respectively).⁽¹⁰⁾ Both static (oxygen-refreshed autoclave) and dynamic (pumped loop) exposures were made at temperatures from 840 to 1290 F. Linear rates of weight loss with time were noted at the higher oxygen level, while much lower weight and dimensional changes were observed at the lower oxygen level. Typical curves are shown in Figure 1. Unlike the Oak Ridge and Westinghouse results, pick-up of interstitials by the vanadium alloys was modest, especially for nitrogen and carbon. Oxygen was the major impurity transferred to the specimens. Titanium as an alloying element caused increased impurity pickup and concentrated the impurities in

a subsurface band 10 to 100 μ thick. The vanadium-chromium binary alloy was the most corrosion resistant and appears to offer promise as a fuel cladding.

Comparison of static and dynamic (20 ft/sec velocity) corrosion rates for oxygen levels of 15 to 20 ppm in the sodium at 1200 F showed that an increase in velocity resulted in an acceleration of the corrosion process for the V-20Ti alloy. However, the corrosion rate was much more sensitive to cold-trap temperature than to sodium velocity. Thus, very low-oxygen sodium may attack vanadium alloys quite slowly, regardless of flow velocity. At Argonne, velocity effects are being looked at separately.⁽¹¹⁾ A disk of vanadium was rotated at different speeds in 600 C (1110 F) sodium containing 16 ppm of oxygen. Periodic weighings of the disk showed that at any given rotational speed, the rate of weight loss was constant, increasing as the square root of the speed above about 50 rpm. This type of dependency suggests that the dissolution is a liquid-diffusion-controlled process.

Inertness of Molybdenum to Sodium

A molybdenum-clad tantalum, control-rod test specimen was exposed by General Electric to 1020 F sodium containing less than 10 ppm oxygen for 1000 hours in a dynamic loop (10 ft/sec velocity) to demonstrate the noble nature of molybdenum under FFTF conditions before in-pile experiments were undertaken.⁽¹²⁾ Following the exposure, visual appearance, gas-pressure leak check, X-ray, ultrasonic, weight check, and dimensional check all indicated that the specimen had undergone absolutely no change.

CORROSION IN RANKINE-CYCLE SPACE POWER SYSTEMS

Lithium-Potassium Systems

A popular concept for nuclear spacecraft propulsion involves an electric generator driven by a potassium-vapor turbine. The potassium is vaporized in a boiler by heat from reactor-heated lithium. The Jet Propulsion Laboratory constructed and tested a two-loop lithium-potassium facility to simulate the major elements of this system.⁽¹³⁾ All material that came in contact with fluids above 1500 F was Cb-1Zr, with the exception of tungsten valve facings, hot-trap getters, and the TZM (Mo-0.5Ti-0.08Zr-0.02C) turbine wheel.

Initial operation of the system for about 1000 hours at lithium temperatures up to 2100 F with a 10 percent bypass yttrium hot trap in the lithium loop and 2200 ppm nitrogen and 700 ppm oxygen in the lithium produced only minor grain-boundary notching of the Cb-1Zr tubing. No evidence of mass transfer of the yttrium was encountered, despite thinning of the yttrium plates in the hot trap and analyses indicating 5 weight percent yttrium in the lithium. Subsequent operation of the system at similar temperatures and times without hot trapping resulted in extensive mass transfer of columbium throughout the system and extensive contamination and corrosion of the Cb-1Zr tubing.

Operation of the potassium loop both with and later without a full-flow zirconium hot trap produced little change in the interstitial content of the system tubing. Corrosion after both modes of operation was minimal. Vapor quality had a marked effect of erosion of the tubing. Low-quality vapor (50 to 80 percent) eroded the outside radii of the tubing and the throat of the turbine nozzle. Higher

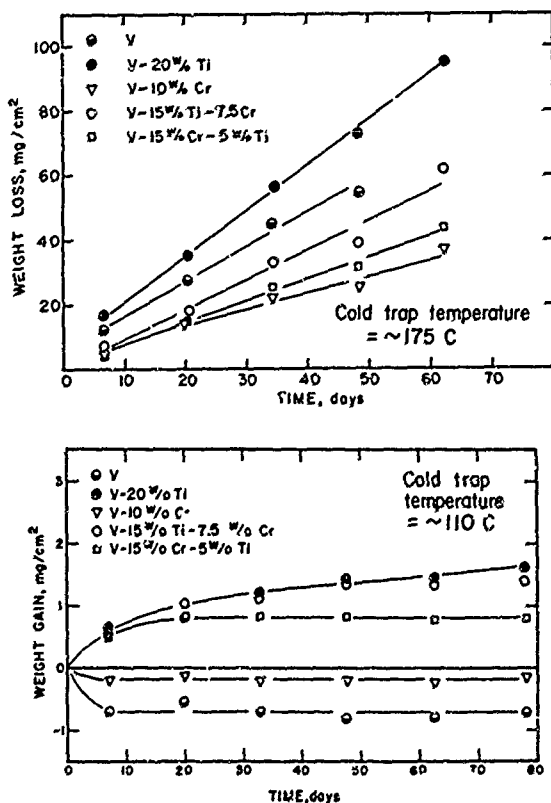


FIGURE 1. CORROSION OF VANADIUM ALLOYS BY COLD-TRAPPED SODIUM IN STATIC OXYGEN-REFRESHED AUTOCLAVES⁽¹⁰⁾

quality vapor had no observable erosive effect, even when run longer and at higher temperatures.

A comprehensive review of the compatibility of structural materials with molten lithium is included in a Lawrence Radiation Laboratory treatise on lithium as a nuclear reactor coolant. (14) It covers virtually all recorded lithium corrosion research using static and dynamic systems, with base metals, refractory metals, ceramics, and cermets. There are special sections on the effects of lithium on mechanical properties of refractory alloys, the resistance of materials to erosion and wear in lithium, and the solubility of containment metals in lithium.

Advanced tantalum alloys for Rankine-cycle space-power systems are being tested by General Electric for resistance to attack by 1885 to 2400 F lithium in thermal-convection capsules. (15) Specimens of T-111 (Ta-8W-2Hf), ASTAR 811C (Ta-8W-1Re-0.7Hf-0.03C), and ASTAR 811CN (Ta-8W-1Re-1Hf-0.015C-0.01N) were exposed for 5000 hours and then bend tested and analyzed chemically and metallographically. Included were some specimens that had been purposely contaminated with oxygen in both the welded and unwelded conditions.

In almost every case, the posttest oxygen content dropped to below 10 ppm (from as much as 387 ppm in oxygen-contaminated T-111, 248 ppm in oxygen-contaminated 811C, and 59 ppm in noncontaminated 811C). The greatest corrosive attack was observed in welded, oxygen-contaminated 811C specimens. Apparently, excess oxygen was free to combine with the lithium to produce platelet-type (intragranular) attack up to 15 mils deep, as well as intergranular attack throughout. Similar specimens given a 2400 F heat treatment for 1 hour to combine all of the oxygen with hafnium prior to lithium exposure were not attacked. On the basis of Cb-12r technology, if there is more oxygen in the alloy than the amount needed to oxidize all of the getter present (hafnium, in the case of these tantalum alloys), these alloys would be attacked by lithium regardless of heat treatment.

Mass transfer of interstitials (carbon, nitrogen, and oxygen) from the hottest portion of the capsule walls to the cooler end was observed. A loss in stress-rupture life of the exposed 811CN specimens tested at 20 ksi and 2400 F was attributed to heat-treatment effects rather than corrosive attack by the lithium.

The observed drastic increase in the apparent solubility of columbium in 1110 F potassium as a result of increasing the oxygen content of the potassium is commonly attributed to the formation of potassium columbates. As an alternative explanation, Klueh analyzed the case where no ternary oxides exist at equilibrium and the higher solubility stems from a reduction in the activity of the columbium by virtue of interactions between the two solutes (oxygen and columbium) in the melt. (16) By applying Wagnerian interaction parameters, he was able to obtain order-of-magnitude agreement with measured oxygen distribution coefficients. Also, by use of the phase rule at equilibrium at 1110 F, he showed that there would be no partitioning of added oxygen, if a third condensed phase (ternary oxide) were present. Since partitioning definitely occurs, then the presence of the ternary oxide is ruled out.

Mercury Systems

Extensive testing of 9Cr-1Mo steel for the tubes of a NaK-heated mercury boiler for the SNAP-

8 Rankine-cycle power-conversion system revealed that this material was not sufficiently resistant to mercury corrosion at 1100 F. In view of the very low solubility of tantalum in mercury, the feasibility of substituting tantalum for the 9Cr-1Mo steel was studied at Aerojet-General. (17) The results of a fabrication and corrosion-test program indicated considerable promise for bimetallic tubes consisting of tantalum liner bonded by hot coextrusion to the ID surface of a Type 316 stainless steel tube. An alternative tube configuration, which also appears acceptable for a 40,000-hour operating life, consists of a circular tantalum tube inside an oval-section tantalum tube with stagnant NaK between them and a circular Type 321 stainless steel tube around the outer one. Mercury flows in the inner tube and NaK flows in the outer annulus.

In another part of the same program, it was noted that mercury-loop contamination by air or Mix 4P3E, an organic fluid used for lubrication and cooling in SNAP-8, resulted in the formation of reaction products on the tantalum tube surface that interfered with heat transfer. A six-step chemical cleaning procedure, with distilled-water flushing after each step, effectively removed the surface film.

SOLUBILITY OF MATERIALS IN LIQUID METALS

Because solubility is a basic parameter which may be important in understanding the mechanism of corrosion of materials by liquid metals, numerous solubility studies are continually being undertaken. Some results which have become available since mid-1969 are presented in Table 2. (18-21)

SOVIET TRANSLATIONS

A 1968 Soviet treatise on corrosion by liquid metals is now available in English translation. (22) It begins with a review of the theories of the various kinds of corrosion by sodium (destruction of surface films, dissolution, intergranular penetration and mass transfer). A chapter is devoted to corrosion of steels and nickel-base alloys in static isothermal systems. This is followed by a chapter on corrosion under dynamic polythermal conditions, including correlations with test results on steels, nickel-base alloys, and refractory metals. A rather extensive discussion of carbon transfer in sodium systems ensues, with Nevegerov's conclusion that carbon migrates as C_2 , which forms as a result of interaction of dissolved oxygen ions in the sodium with carbon ions in the steel. The final chapter reviews various physico-chemical mechanisms of corrosion and carbon transfer in sodium and other liquid metals, with emphasis on the importance of the ionic nature of nonmetallic impurities in the melts. This book represents a good summary of Soviet corrosion research over the past decade, and the author's approach to the general theory of corrosion by liquid metals is noteworthy.

A 437-page translation of a Soviet monograph on thermal properties of metals contains brief chapters on heat capacity and thermal conductivity of liquid metals. (23) The author, Filippov, reviews the literature covering both theoretical and experimental studies of these properties, makes some guarded observations, and concludes that further experimental studies are needed before positive correlations can be made. He notes that the change in heat capacity on melting is generally small (3 percent for tin, 3.5 percent for lithium, 1.5 percent for sodium, 6 percent for lead, and negligible for mercury). Further, he observes that curves of heat capacity of

all of the above liquid metals against the ratio of absolute temperature (T) to absolute melting point (T_{mp}) fall within a narrow band. For example, for sodium, potassium, and lithium (Figure 2), the maximum deviation from the average (dotted) curve is 3 percent. This suggests that it may be possible to infer heat capacities from T/T_{mp} for a large number of liquid metals.

Literature values for thermal conductivity (k) of molten tin, lead, cadmium, bismuth, lithium, sodium, and potassium generally show wide scatter and are not amenable to treatment as a generalized function of temperature. The ratio k_{solid}/k_{liquid} at the melting point of many metals is very close to the ratio of electrical conductivities, the latter ratio usually being slightly greater. The electronic component of heat conduction follows the Wiedemann-Franz law near the melting point, but deviates increasingly with rising temperature.

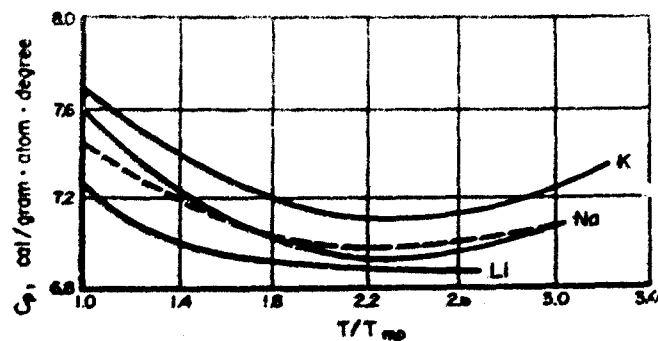


FIGURE 2. SIMILARITY OF DEPENDENCE OF HEAT CAPACITY ON T/T_{mp} FOR THREE ALKALI METALS(23)

TABLE 2. SOLUBILITY OF VARIOUS MATERIALS IN LIQUID METAL

Solvent	Solute	Solubility Equation	Temperature Range, K	Heat of Solution, kcal/mole	Reference
K	Mo	$\log(\text{wppm Mo}) = 2.21 \pm 0.1 - (1472 \pm 156)/T(K)$	1040-1316	6.7 ± 0.8	18
K	W	$\log(\text{wppm W}) = 5.02 \pm 0.21 - (3851 \pm 243)/T(K)$	1055-1328	17.6 ± 1.2	18
Pb-55.5 Bi	Cu	$\log(\text{wppm Cu}) = 6.41 - 1920/T(K)$	673-823	8.78 ± 0.27	19
Pb-55.5 Bi	Co	$\log(\text{wppm Co}) = 5.30 - 2834/T(K)$	673-823	12.97 ± 0.50	19
Pb-55.5 Bi	Ni	$\log(\text{wppm Ni}) = 5.70 - 1000/T(K)$	753-823	4.57 ± 0.53	19
Pb-55.5 Bi	Cr	$\log(\text{wppm Cr}) = 3.98 - 2280/T(K)$	673-823	10.43 ± 2.28	19
Na	Cu	$\log(\text{wppm Cu}) = 5.6362 - 3199.5/T(K)$	595-999	14.6 ± 0.7	20
Na	Bi	$\log(\text{at. \% Bi}) = 1.188 - 2100/T(K)$	398-563	9.6	21
Na	Bi	$\log(\text{at. \% Bi}) = 4.900 - 4178/T(K)$	563-923	19.1	21
Na	Mo	< 2 wppm Mo (insufficient analytical sensitivity)	873-1278	---	21
Na	Ti	1 to 16 wppm (no regular temperature dependence)	873-1173	---	21
Na	Cr	1 to 200 wppm (large scatter)	873-1273	---	21

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